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Laser ablated pure non-crystalline Co thin films for inductors for ultra-high frequencies

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ABSTRACT

Non-crystalline Co thin films have been prepared by pulsed laser ablation deposition. From the M-H hysteresis loops measurements, a soft magnetic behavior is observed. Néel type magnetic domain walls are observed in the as-deposited films. The spontaneous magnetization, M_s ($T = 300$ K), is ≈ 860 emu/cm³. After annealing at 500 °C, M_s ($T = 300$ K) is ≈ 1460 emu/cm³. The extrapolated to zero K resistance decreases almost two orders of magnitude from the as deposited samples to the crystallized heated at 500 °C ones. A trilayer Co/Cu/Co has shown a real part magnetic susceptibility of 120 at 100 MHz. In the 100 MHz to 1 GHz frequency range, a perpendicular bias magnetic field increased this value up to 270, remaining almost constant for all range.

INTRODUCTION

Ferromagnetic elements in the amorphous state have attracted interest from both the theoretical and the applied points of view [1-9]. Different techniques have shown the possibility of producing these kind of materials; in particular, the pulsed laser ablation deposition method, PLAD, is one of the most versatile method for preparing non-crystalline solids [7] and thin films of pure ferromagnetic elements at room temperature [8-9], exhibiting interesting magnetic and electrical properties.

Since years, there has been interest in developing thin film soft ferromagnetic materials appropriated for high frequency ($f > 100$ MHz) applications, as for example, for electromagnetic devices as a high frequency field-amplifying component, e.g., in read-write heads for magnetic disk memories for computers [10-12]. Some of the most desirable properties in these materials are high saturation magnetization, low magnetic coercivity, high magnetic permeability and high electrical resistivity.

In this work we show that non-crystalline PLAD Co thin films - films made up of one element only -, are very suitable for high frequencies applications because they exhibit, at room temperature, such magnetic and electrical properties mentioned above.

EXPERIMENTAL

Non-crystalline Co thin films were prepared by PLAD, using a stainless-steel chamber (Neocera) at 10^{-5} mbar pressure. A pulsed Nd: YAG laser, $\lambda = 1024$ nm, 20 Hz repetition rate, with 4 ns pulses, energy 300 mJ per pulse, was used. A polished circular disk, 20 mm in diameter, of pure Co (Goodfellow metals, 99.999%) was used as the target, which was rotated at 32 rpm. The area of the laser beam on the target was 2 mm². Pure Si (111) substrates, 2 x 25

mm², were used for depositing the CoCuCo sandwiches and 4 x 4 mm² substrates were used for the pure Co films deposition. The substrates, which were rotated at 120 rpm, were situated at 70 mm from the target. The deposition time was 30 min., thus obtaining samples \approx 250 nm thick. After deposition, some samples were annealed at temperatures up to 700 °C for 10 minutes in an inert atmosphere.

Trilayers Co/Cu/Co were also fabricated by this technique. For this purpose, two different targets were placed into the chamber: one of Co and one of Cu. The Co layers were deposited in accordance with the above described procedures. The intermediate Cu layer was deposited using a circular pure Cu (Goodfellow metals, 99.999%) target, 20 mm in diameter, that was also rotated at 32 rpm. The deposition time was 120 min., thus obtaining a layer 0.5 μ m thick.

X- ray analysis were performed with a diffractometer (XRD-3000 Seifert), using Mo anode and a grazing incidence geometry. A graphite secondary monochromator was placed just before the detector.

We performed different magnetic measurements. We measured the Magneto-optical Transverse Kerr Effect, MOKE. The diameter of the light spot was 0.5 mm. An EG&G Vibrating Sample Magnetometer, VSM, was used for measurements at room temperature and in magnetic fields up to 1 Tesla, and a Superconducting Quantum Interference Device, SQUID, was used for measuring in magnetic fields up to 5 Tesla.

The Bitter technique was applied in analyzing magnetic domain configurations. Improved contrast and pattern resolution were achieved by applying a magnetic field perpendicular to the sample's plane, in order to polarize the colloidal suspension.

The electrical resistivity, ρ , of the Co thin films was measured by a conventional four probes system. The dependence of ρ on temperature was studied from room temperature up to 500 °C. These measurements were performed in a furnace where an Ar atmosphere was kept to avoid oxidation processes in the samples.

The susceptibility spectra of the films was measured both with a LCZ meter (Keithley 3322) in the frequency range from 100 Hz to 100 kHz and also with a network analyzer (HP 8753D) in the range from 100 kHz to 1 GHz. In the low frequency range (up to 100 kHz) the susceptibility of our films was determined by measuring the change of inductance of a pick up coil, when the sample was inserted in it and after applying an external bias magnetic field. In the high frequency range, the magnetic susceptibility was measured using a strip-loop device – 40 mm long, 8 mm wide and 3 mm high - mounted on a standard SMA connector. The changes of the impedance of the fixture when the sample was inserted in it were measured with the network analyzer. In both situations, at low and at high frequencies, an external bias magnetic field, created by a pair of Helmholtz coils, was applied in the plane of the sample and perpendicular to the ac magnetic field inside the pick up coil or inside the strip-loop fixture.

RESULTS AND DISCUSSION

Microstructure

X-ray diffractograms were performed on the as-deposited as well as on the annealed Co thin films. For the as-deposited samples, the analysis showed none of the peaks characteristic of crystalline structures, see figure 1(a). However, for the samples heat treated at 500 °C in an Ar atmosphere for 10 min., a superposition of both Co fcc and Co hcp crystalline diffraction pattern

was observed exclusively, as shown in figure 1(b); only the first peak centered at ≈ 17 deg. in figure 1(b) could be due to some Co oxide present in the sample.

Magnetic and electrical properties

The MOKE measurements revealed a bistable soft magnetism. Figure 2 shows two MOKE hysteresis loops, M-H, obtained when an in plane magnetic field is applied along the longitudinal easy direction and perpendicular to the longitudinal direction respectively: an easy coercive field of 1.7 Oe and a weak in plane magnetic anisotropy, anisotropy field of 12 Oe, were measured.

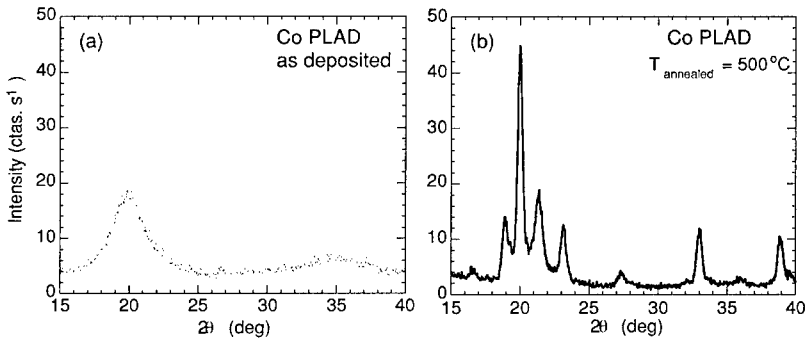


Figure 1. X-ray diffractograms from the Co thin films: (a) as-deposited: none of the peaks characteristic of crystalline structure are observed; (b): after 10 min. at 500 °C. A superposition of both Co fcc and Co hcp crystalline diffraction pattern was observed exclusively; only the first peak centered at ≈ 17 deg. could be due to some Co oxide.

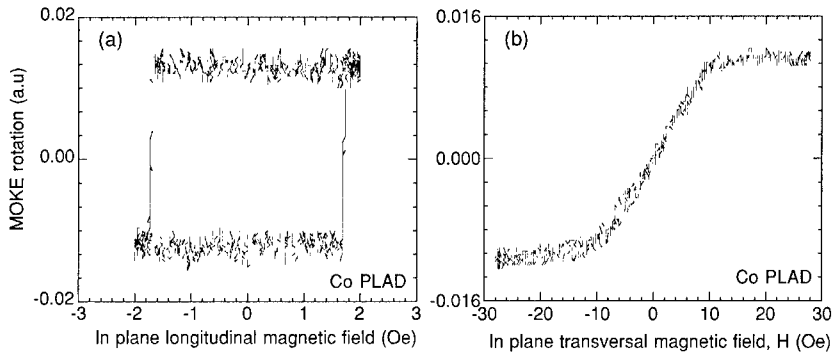


Figure 2. MOKE hysteresis loops, M-H, obtained when an in plane magnetic field is applied along (a) the longitudinal easy direction and (b) perpendicular to this easy longitudinal direction.

From VSM measurements, it is confirmed that for the as-deposited sample the coercive field is ≈ 1.7 Oe, as it is shown in figure 3(a). The inset in this figure shows the measurements of M-H performed in a SQUID in fields up to 5 Tesla. It is seen that the spontaneous magnetization, M_s , at room temperature is 860 emu/cm^3 . When the samples are annealed at 500°C , M_s increases to 1460 emu/cm^3 , see figure 3(b). This M_s value corresponds to the value of M_s , at room temperature, of pure crystalline Co.

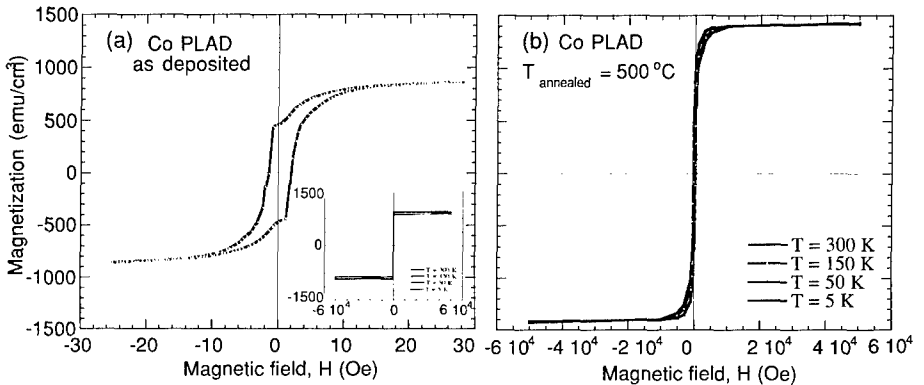


Figure 3. (a) VSM measurement for the as-deposited sample. The coercive field is ≈ 1.7 Oe and $M_s(T = 300 \text{ K}) = 860 \text{ emu/cm}^3$. The inset shows the SQUID measurements at 5 Tesla.; (b) SQUID measurements for the sample annealed at 500°C , $M_s(T = 300 \text{ K}) = 1460 \text{ emu/cm}^3$. Note that the vertical scale is the same in the two graphics (a) and (b).

The magnetic domains structure corresponding to the as-deposited Co thin films are shown in figure 4(a,b). The Bitter patterns show Néel type magnetic domain walls. This irregular shape of the domain walls can be attributed to a weak in plane anisotropy and to a local deviation of the magnetization from its average orientation. Figure 4(c,d) is a schematic representation of the structures observed in figure 4(a,b), indicating the direction of M_s , inside each domain.

Figure 5 shows the dependence of the electrical resistivity on temperature, $R(T)$. Note that the resistance decreases almost two orders of magnitude from the as-deposited sample to the sample heat treated at 500°C . It can be also seen two irreversible jumps in the resistance, associated with the irreversible changes in the microstructure of the films occurring when they are annealed.

Inductive behavior

At low frequency, $f = 100 \text{ Hz}$, the value of the magnetic susceptibility of square 4 mm long Co samples was 1200. As the frequency of the ac magnetic film increases, the motion of the magnetic domain walls is hindered, and consequently the magnetic susceptibility decreases to 200 at 100 kHz . In this situation, see figure 6(a), the application of a perpendicular bias magnetic field contributes to create a preferential orientation of the magnetization in the Co film,

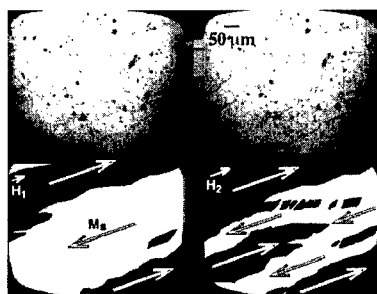


Figure 4.(a,b) Magnetic domains for the as-deposited films; a magnetic field changes the domains structure; (c, d) schematic picture of both photos.

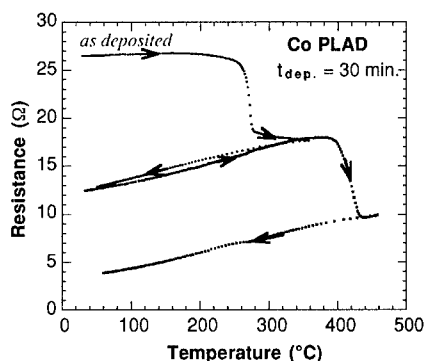


Figure 5. Dependence of the resistance on temperature of Co PLAD thin films. The arrows indicate the direction in which the measurements were done.

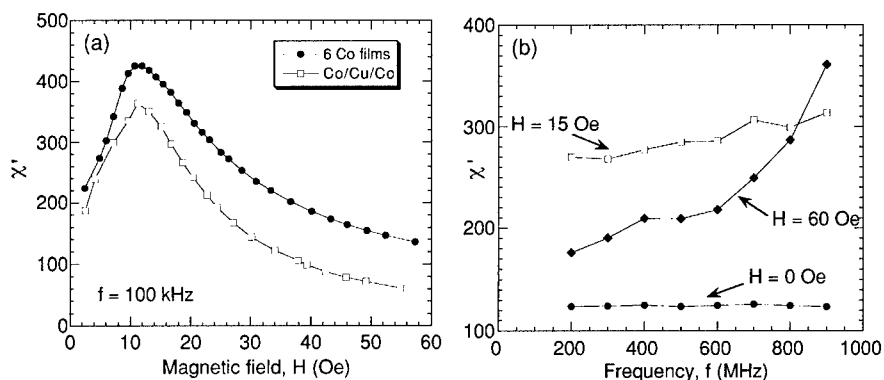


Figure 6.(a). Evolution of χ' with a dc magnetic field at $f = 100$ kHz, corresponding to two different systems: six Co films and a trilayer Co/Cu/Co, respectively; (b) high frequency behavior of the trilayer for different bias magnetic field.

along the direction of this bias field. In this situation, the ac magnetic film produces a coherent spin rotation towards the direction of the ac field, and this rotation mechanism gives rise to an increase of the magnetic susceptibility, up to 450 in our Co films. Further increments of the bias field tend to fix the magnetization along the dc magnetic field direction and the susceptibility decreases again.

The susceptibility was measured up to 1 GHz, see figure 6(b). The real part of the susceptibility, χ' , shows a slight frequency dependence, which could be due either to a measuring artifact as the copper strip-loop exhibits a resonance at 1.3 GHz, and also to the effect of ferromagnetic resonance of Co films. χ' of our Co films is of the same order than χ' in other magnetic films, i.e. $\chi' \approx 1000$, for Co-Zr [10], $\chi' \approx 500$ for Co-Nb-Zr [11], or $\chi' \approx 200$ for Fe-Cr-Ta-N [12].

CONCLUSIONS

PLAD non-crystalline Co thin films showed a soft magnetic behavior: an easy coercive field of 1.7 Oe and an in plane magnetic anisotropy, ≈ 12 Oe, were exhibited. The value of the spontaneous magnetization at room temperature was, M_s ($T = 300$ K) ≈ 860 emu/cm³. After annealing at 500 °C, M_s ($T = 300$ K) ≈ 1460 emu/cm³. The extrapolated to zero K resistance decreased almost two orders of magnitude from the as deposited samples to the heat treated at 500 °C. Due to their soft magnetic properties, high M_s , high values of the magnetic susceptibility at high frequencies and high electrical resistivity, these films were suitable for ultra high frequencies applications. The magnetic susceptibility was 200 at 100 kHz. The application of a perpendicular bias magnetic field increased the magnetic susceptibility up to 450. In the 100 MHz to 1 GHz frequency range, the real part of the susceptibility remained almost constant and its value was similar to those of the Co-Zr, Co-Nb-Zr and Fe-Cr-Ta-N compounds.

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